鉄合金のマルテンサイト・ベイナイト変態における不均一核生成の結晶学

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作成例＜電子データ（CD-R）にて提出するもの＞

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論文内容要約

Chapter 1: Introduction

Grain refinement of lath martensite or bainite structures in structural steels are still demanded for further strength and toughness. Since controlling nucleation site is useful for grain refinement, inclusions, precipitates and dislocations have been introduced as nucleation sites of martensite and bainite transformation by thermal or thermomechanical processes. However, it is known that specific variants of α' martensite among 24 K-S possible ones in K-S orientation relationship are formed preferentially during those processes. For example, it is reported that coarse bainite formed at higher temperature by strong variant selection, resulting in loss of toughness. Although variant analysis needs orientation of prior austenite, lath martensite and bainite do not have enough amount of austenite in those structures. Thus, variant selection rule and formation mechanism of martensite and bainite structure is not clarified yet when martensite and bainite transform from work hardened or precipitated austenite. In this study, we investigate the variant selection appearing in grain refinement process by using reconstruction of austenite orientation method.

Chapter 2: Experimental procedure

In this Chapter, the experimental procedure which is used in this thesis commonly is explained.

Chapter 3: Variant selection by deformation prior to transformation in lath martensite

Ausforming is an effective thermomechanical process to develop the strength without loss of toughness in martensite or bainite structure. Martensite or bainite transform from work hardened austenite in this process. However, it is reported that variant selection occurs during ausforming, and the selection rule is not clarified yet. In this chapter, variant selection by ausforming has been investigated in uniaxially compressed or non-compressed Fe-18Ni-xC (x = 0, 0.1 and 0.3) (mass %) alloys which forms lath martensite.

Table 1 Various factor of variant selection in ausformed martensite.

Factor		Lath M	Lenticular M poly	Lenticular M single
Stress effects	Residual stress	○ (Weak)	○ (Weak)	○
	Stress field around dislocations	×	×	×
Interfacial boundary effects	Microband in γ (low angle boundary)	○	×	×
	γ grain boundary (high angle boundary)	○	○	—

It is revealed that some variants whose habit planes (HPs) are parallel to active slip plane in austenite are selected preferentially in ausformed lath martensite. Microbands are developed along to active slip plane in austenite of 50% compressed Fe-25Ni-0.5C alloy in which is austenite at room temperature. The variant selection model considering nucleation at dislocations in austenite cannot explain the selection rule in ausformed lath martensite. The growth directions or HPs of selected variants are nearly parallel to austenite grain boundaries or microbands, respectively. Thus, selected variants are consistent with variants expected by nucleation models at austenite boundary and at microbands model. In addition, it is confirmed a presence of compressive stress with approximately 100MPa along to compression axis in 50% compressed austenitic Fe-25Ni-0.5C alloy. Selected variants can be explained by the residual stress accelerating lattice change from fcc to bcc by partial dislocation. Variant selection models are summarized in Table 1.

Chapter 4: Variant selection by deformation prior to transformation in lenticular martensite

In this chapter, variant selection by ausforming has been investigated in lenticular martensite in compressed or non-compressed Fe-25Ni-0.5C and Fe-32Ni (mass %) alloys. Furthermore, to investigate effects of austenite grain boundary, single crystalline specimen is made from Fe-32Ni alloy.

It is revealed that the variants whose HPs are perpendicular to active slip plane in single crystalline specimen are selected by ausforming. On the other hand, variants whose HPs are parallel to compression plane are selected in ausformed poly-crystalline specimen. Therefore, it is revealed that variant selection differs due to the presence of austenite grain boundary in lenticular martensite, and besides, different variant selection occurs comparing with lath martensite in poly-crystalline specimen. In lenticular martensite, selected variants cannot be explained by nucleation at dislocation model and nucleation at microbands model in both specimen. However, selected variants in single crystalline specimen can be explained by selection models considering effects of residual stress on the two partial dislocation assuming lattice change from fcc to bcc. On the other hand, selected variants in poly-crystalline specimen can be explained by preferential nucleation and growth models along austenite grain boundary. Variant selection models are summarized in Table 1.

Chapter 5: Grain refinement of lath martensite by displacive cyclic transformation in Fe-high Ni alloys

In this chapter, we apply the cyclic transformation to Fe-18Ni, Fe-18Ni-0.1C and Fe-23Ni alloys, in which displacive austenite transformation appears during heating, in order to accumulate dislocation through displacive forward and reverse transformations for further grain refinement.

Fig. 1 summarizes the average intercept length of high angle grain boundaries of all specimens as a function of number of cycles. It was found that one cycle process of forward and reverse transformations effectively refines martensite structure in Fe-18Ni, Fe-18Ni-0.1C and Fe-23Ni alloys. In particular, martensite structure in Fe-18Ni-0.1C alloy were further refined with increasing cycles unlike in the case of carbon free alloys. This further refinement in Fe-18Ni-0.1C alloy is due to acceleration of accumulation dislocation and formation of fine austenite grain by recrystallization during cyclic treatment.

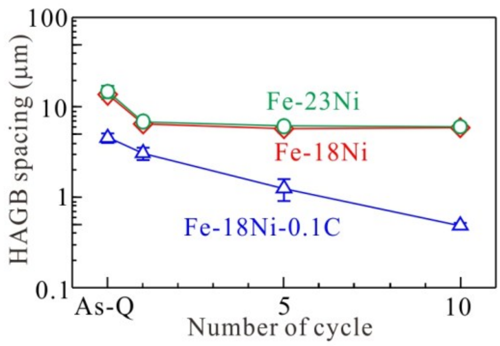

Fig. 1 Average intercept length of high angle grain boundaries against number of cycles.

Chapter 6: Effects of MnS and VC on martensite/bainite transformation in Fe-Mn-C alloys

In this chapter, effects of single or complex precipitation of MnS and VC on the martensite and bainite structure are investigated in Fe-2Mn-0.2C alloys for the further refinement.

Fig. 2 (a) and (b) show optical microscopic images in early stage of bainite transformation at 773K in Base and VC alloy, respectively. While coarse bainite structure is formed at austenite grain boundary in the Base alloy (Fig. 1(a, c)), bainite structure formed at austenite grain boundary is significantly refined and suppressed in

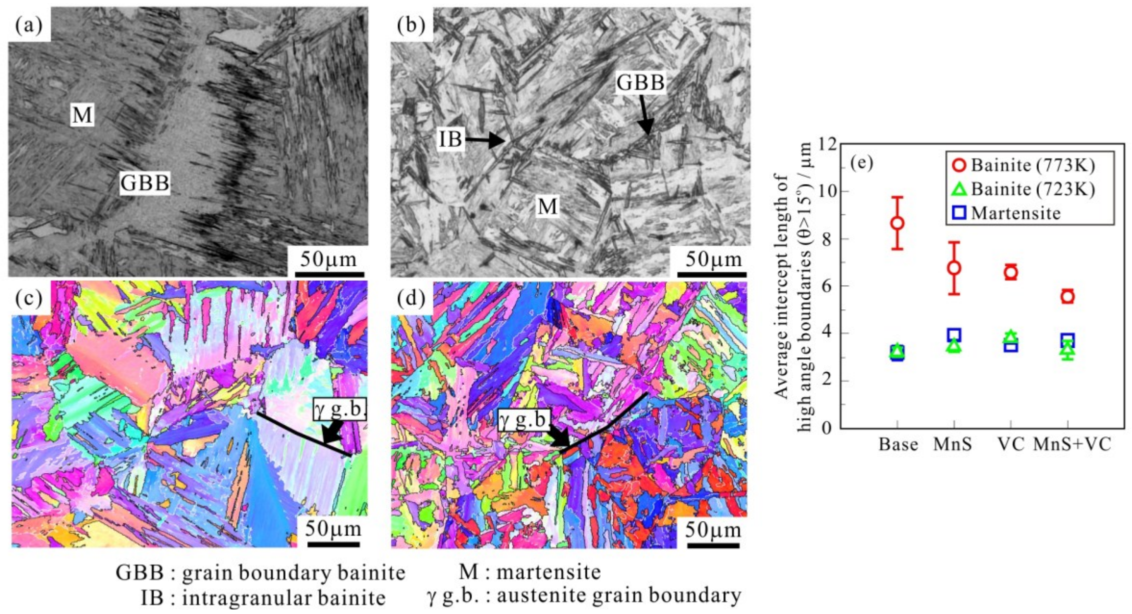

Fig. 2 Optical microscopic images of bainite structure transformed at 773K for 10s in (a) Base and (b) VC alloy, and α orientation maps of bainite transformed completely in (c) Base and (d) VC alloy. (e) Average intercept length of high angle boundaries (θ > 15°).

the VC alloy (Fig. 1(b)). Furthermore, intragranular bainite transformation is observed in VC alloy (Fig. 1(b)). Fig. 2 (c) and (d) show α orientation maps of completion of bainite transformation at 773K in Base and VC alloy, respectively. In VC alloy, bainite structure becomes finer compared with Base alloy. EBSD analyses reveal that multiple variants are formed at austenite grain boundary in VC alloy because VC precipitated at austenite grain boundary can play as the nucleation site of bainite transformation in addition to austenite grain boundary while single variant forms dominantly at austenite grain boundary in Base alloy. Fig. 2 (e) shows the average intercept length of high angle grain boundaries of which misorientation (θ) is larger than 15^{o}. In transformation at 773K, bainite structure becomes finer by adding MnS or VC and the combined addition of MnS and VC results in the finest structure. On the other hand, little effect of MnS or VC on grain size is found in bainite transformed at 723K and as-quenched martensite.

Chapter 7: Conclusions

In this chapter, each chapter through 1to 6 are summarized.